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**CORROSION RESISTANCE OF CONSTRUCTIONAL**

**MATERIALS IN ALKALI METALS**

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Nowadays the heat-transfer metals has been broadly used for nuclear power plants both in the USSR and in other countries.

As a result of the work performed, it was shown that the most perspective coolants are the alkali metals, such as sodium, sodium-potassium alloy and lithium. The study of these coolants required especial selection and development of the constructional materials which were corrosion-resistant in these mediums, and also to develop the methods of tests and investigations.

The present work is devoted to the scientific and engineering problems of corrosion resistance of materials in sodium and lithium.

**INVESTIGATIONS ON CORROSION RESISTANCE OF  
CONSTRUCTIONAL MATERIALS IN SODIUM**

**1. Chemical Nature of Oxygen Dissolved in  
Sodium Coolant**

In many works it was shown that corrosion rate of constructional materials to a considerable degree depends upon the oxygen content in sodium. Speaking about the oxygen dissolved in sodium, all the authors mean its molecular form  $\text{Na}_2\text{O}$ . However, Bowman and Cubicciotti [1] suggested that  $\text{Na}_2\text{O}$  dissolved molecules were dissociated in sodium into ions. Irrespective of the statement of these authors we came to a conclusion that the partial dissociation of  $\text{Na}_2\text{O}$  was possible, this was confirmed by experiments [2, 3, 4]. The study of oxygen electric transfer in sodium at  $300^\circ$  and  $600^\circ\text{C}$  showed that oxygen was transferred only to the positive pole of a cell, the quantity of transferred oxygen being practically constant for

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1 a.h. passed direct current (3600 coulombs). The electric transfer of its chemical analogue, i.e. sulphur, was also discovered. This experiment is very obvious in a glass cell because sodium sulfides are of brown colour. Proceeding from these results we assume that there is  $\text{Na}_2\text{O} \rightleftharpoons 2\text{Na}^{+1}\text{O}^{-2}$  equilibrium between the molecular and ionic forms of oxygen in liquid sodium. In our opinion, the basic reason for the dissociation of  $\text{Na}_2\text{O}$  polar molecules into ions is the orientational interaction of sodium atoms with the polar ends of  $\text{Na}_2\text{O}$  molecules, as a result of which (owing to the deformation of orbits of valent electrons) the atoms of sodium gain induced dipoles. As a result, the polar ends of  $\text{Na}_2\text{O}$  molecules locally can be solvated by sodium atoms with induced dipoles, and the further process is similar to the dissociation of electrolytes. Apparently a portion of  $\text{Na}_2\text{O}$  molecules may be in excited state even with weak bonds.

All these data make possible to draw an important conclusion that having been dissolved in sodium, oxygen may have different thermodynamic activity right up to unity when it is in ionic form. This chemical nature of oxygen dissolved in liquid sodium can undoubtedly cause a significant accelerating effect upon the corrosion of metals and to determine its mechanism.

## 2. Study of Corrosion of Constructional Materials in Flowing Sodium

The corrosion resistance of a great number of constructional steels was studied after their tests in pump loop with 5 metres per sec. sodium flow at 500, 600 and  $700^{\circ}\text{C}$ . The tests were carried out for 4 000 to 6 000 hours. In order to purify sodium from oxygen a cold trap which maintained the oxygen content in the range of  $3 + 8 \cdot 10^{-3}$  weight per cent was used. After the tests the steel samples were investigated.

The investigations showed that at 500 and  $600^{\circ}\text{C}$  low-alloy perlite steels (C from 0.11 to 0.17 per cent, Cr from 1.1 to 2.19 per cent, Mo from 0.25 to 1.54 per cent, V from 0.20 to 0.32 per cent) became considerably decarbonized in this connection, their strength decreased by 25-30 per cent and their ductibility increased by 10-20 per cent. The steels of 5 to 12 per cent chrome content did not undergo the decarbonization.

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In static conditions at 500°C the same perlite steels do not become decarbonized in sodium. In work (4) it is shown that transfer rate depends upon  $\Delta\text{CO}$ , i.e. oxygen content and the transfer proceeds by way of formation of  $\text{CO}$ , i.e. carbon monoxide. In dynamic conditions at the sodium rate of 5 meters per sec. steel samples are supplied with such quantity of oxygen that the decarbonization reaction, may proceed.

Under the same conditions, but at the temperatures of 600 and 700°C, the samples of austenitic steels (C from 0.03 to 0.10 per cent, Cr - from 14 to 21.0 per cent, Ni from 8 to 14 per cent with the additions of Ti, Nb and Mo from 0.5 to 1.5 per cent) were tested. When tested during 5100 hours at 600°C some steels, such as: Cr = 17.0 per cent and Ni = 12 per cent; Cr = 18 per cent and Ni = 9 per cent ; Cr = 20 per cent and Ni = 14 per cent underwent intercrystalline corrosion to the depth of 150 $\mu$ , 100 $\mu$  and 70 $\mu$  respectively. At the temperature of 700°C the same steels underwent corrosion to the depth not exceeding 30 $\mu$ . The analysis of pump loop operation showed that during the whole period of tests oxygen content in sodium was equal to  $6 + 8 \cdot 10^{-3}$  per cent and  $3 + 5 \cdot 10^{-3}$  per cent at 600°C and 700°C, respectively. Thus, it is quite probable that in this case the corrosion rate of austenitic steels tested at 600°C turned out to be considerably greater than that at 700°C, owing to the increased content of oxygen in sodium. The results given in work (5) are indicative of the probability of this explanation.

This experiment (carried out in 5 meters per sec. sodium flow) clearly shows that in a circulating non-isothermal system it is important to maintain oxygen content not more than  $5 \cdot 10^{-3}$  per cent in order to ensure the high corrosion resistance of austenitic steels. At the temperature of 700°C the steels containing silicon and titanium in the quantities of 2.6 per cent and up to 1.0 per cent, respectively, considerably changed their mechanical properties. For example, the strength of steel containing silicon 2.6 per cent increased from 46 kg/mm<sup>2</sup> (after hardening from 1050°C) to 60 kg/mm<sup>2</sup>, and its relative elongation decreased from 32.0 to 5.0. X-ray structural investigation of samples of these steels showed that the changes of mechanical properties were explained by the formation of surplus phases such as:

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sigma-phase and secondary carbides. These phases were found in the samples of austenitic steels tested at 700°C to both in sodium flow and in argon (without any medium).

For the work at higher temperatures with the use of sodium coolant the materials of high heat resistance are required. Therefore the materials such as: nickel-base alloys (Ni from 60 to 75 per cent; Cr from 16 to 20 per cent; Mo up to 5 per cent; Ti from 1.0 to 2.0 per cent; W up to 10.0 per cent) and for comparison the austenitic steel containing chrome 20 per cent and nickel 14 per cent, were chosen for the test. The corrosion tests of samples of nickel-base alloys and of steel were carried out in a pump loop with 2 meters per sec. sodium flow at the temperatures of 700, 750, 800, 850, 900 and 930°C. At each of these temperatures the time of test was equal to 300 hours. In all cases the temperature of electromagnetic pump was 200-220°C. The cold trap provided oxygen content in sodium not over  $5 \cdot 10^{-3}$  per cent.

As a result of the tests, it was revealed that nickel-base alloys suffered considerable intercrystalline corrosion but the austenitic steel practically was not corroded. According to the data of weight losses by samples (corrosion rate in  $\text{mg}/\text{sm}^2\text{-hour}$ ), the diagram characterizing the tendency of alloys and steel to the mass transfer in sodium depending on the temperature was built (Fig. 1). This diagram shows that at 700-750°C all the materials are of the same high corrosion resistance. With the further rise of temperature the materials show already different tendency to the mass transfer depending on their chemical composition. The alloys containing nickel 75 and 60 per cent show the greatest tendency to the mass transfer but at the same time 30 per cent addition of aluminium noticeably decreases this tendency.

The refractory metals, such as molybdenum and niobium can be used for the work in sodium coolant at high temperatures. Molybdenum and niobium in form of plate samples were tested in 5 meters per sec. sodium flow at 600°C during 3 600 hours. During the test oxygen content varied in the range of  $2 + 4 \cdot 10^{-3}$  per cent. As a result the niobium samples became completely destroyed, while molybdenum practically did not suffer any corrosion. The tests of molybdenum were also carried out at 900°C with 2 meters per sec. sodium flow at different oxygen content

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ranging from  $2 \cdot 10^{-3}$  to  $1 \cdot 10^{-2}$  per cent during 210 to 360 hours. Metallographic investigation showed the complete absence of corrosion interaction of molybdenum samples with sodium at any concentrations of oxygen. All samples had very small increase of their weights which practically was constant at different oxygen contents (Table 1).

The above-mentioned results show that molybdenum in sodium coolant is of high corrosion resistance in a broad range of oxygen content. According to the published data the corrosion rate of niobium in sodium depends on oxygen content and becomes very small only with the oxygen content of some  $10^{-4}$  per cent (6).

It was of interest to investigate the dependence of niobium corrosion resistance on oxygen content in sodium coolant. For this purpose the plate samples of niobium were tested in pump loop with 1 meter per sec. sodium flow during 300 hours at  $900^{\circ}\text{C}$  with the oxygen content of  $1 \cdot 10^{-3} - 2 \cdot 10^{-3}$  (with the use of a hot trap),  $1 \cdot 10^{-2}$  and  $4 \cdot 10^{-2}$  per cent. Metallographic investigation showed that at all concentrations of oxygen the samples of niobium suffered uniform corrosion with some weight loss. Uniform corrosion rates (in  $\text{mg}/\text{sm}^2\text{-hour}$ ) are shown in Fig. 2. Thus, from the above-mentioned results one can see that niobium corrosion rate at  $900^{\circ}\text{C}$  to a great extent depends on oxygen content in sodium coolant.

#### CORROSION RESISTANCE OF CONSTRUCTIONAL MATERIALS IN LITHIUM

##### a) Mechanism of Interaction

Chrome stainless steels are the most corrosion resistant in lithium. The chrome-nickel steels substantially lose their strength and ductility above  $550-600^{\circ}\text{C}$ .<sup>5</sup>

The investigations showed that the corrosion processes were the most intensive in the surface layers which were in contact with lithium and first of all, along the bounds of grains. Therefore when studying the mechanism of corrosion the method of layer-by-layer analysis was used.

The study of these layers showed that the most significant process determining the corrosion effect of lithium upon the stainless chrome steels was the decrease of carbon and

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chrome content in the surface layer. The decarbonization of surface layer is connected with the high chemical affinity of lithium with carbon, and its chrome impoverishment depends on the noticeable dissolubility of chrome in lithium particularly above 700-800°C (7). Under isothermal conditions of test a corrosion layer becomes impoverished with chrome to a greater degree when the materials of sample and container are dissimilar, as well as at higher temperatures (Table II). When testing the chrome steels in containers made of armco iron, chrome is transferred on to the wall of a container, and chrome content increases in it up to 1.5 - 2.5 per cent. When the materials of sample and ampule are the same, liquid lithium quickly becomes saturated with chrome, and the system comes to the near equilibrium state.

During static tests in isothermal conditions the corrosion process proceeds only during the first 10-100 hours, and afterwards stops, therefore the increase of duration of tests up to 1 500 - 3 000 hours does not noticeably effect upon the structure and mechanical properties of chrome stainless steels.

The nature of corrosion process established for the simple chrome steels (0X13, IXI3, 2XI3) also remains the same for the complex alloy steels, e.g. IXI2M2 BΦ . In this steel during the first 50 hours the rate of decreasing the chrome concentration in the surface layer is lower (as compared with that in steel IXI3) approximately by one order, but during longer tests they become nearer (Fig. 4a, b).

The decrease of carbon and chrome content substantially changes the structure; i.e. the quantity of carbide phase decreases, microhardness is reduced, and the bounds of grains are noticeably destroyed on the surface.

The same mechanism of corrosion process in lithium also takes place for chrome-nickel steels, however, in this case carbon, chrome and nickel from the surface layer simultaneously become dissolved in lithium. The relative decrease of nickel concentration in the surface layer is higher than that of chrome concentration, because nickel is of greater dissolvability in lithium.

When studying the corrosion processes the radioactive isotope carbon C<sup>14</sup> was also used. Integral radioactivity on the  
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surface was assumed to be proportional to the concentration of carbon on this surface. Isotope C<sup>14</sup> (introduced into lithium) practically did not change the carbon composition of lithium.

In work (8) it was shown that when testing iron in lithium at the temperatures corresponding to -and -modifications iron did not dissolve carbon. On the contrary, when testing carbon steels in lithium all carbon diffuses into lithium, and the structure of these steels becomes purely ferritic. When testing chrome, as well as steels containing chrome the transfer of carbon from lithium on these materials takes place. According to the distribution of radioactive carbon in the surface layer, it is possible to indirectly judge of the distribution of chrome. This assumption is confirmed by Fig. 3 which gives the curves of layer-by-layer distribution of chrome, integral activity of isotope C<sup>14</sup> radiation, and change of microhardness. In this case the proportionality condition between carbon concentration and chrome concentration is valid only for the surface layer which becomes impoverished with chrome. If chrome content reaches the initial one, carbon concentration becomes maximum, and afterwards the diffusion of carbon into deeper layers which have not undergone any changes with respect to chrome content is observed. From the depth of layer with maximum chrome content one can judge of the thickness of layer, where impoverishment with chrome proceeds actively, and from the degree of decreasing the radioactive radiation on the surface one can judge of the decrease of chrome concentration.

The micrographs and auto-radiograms of radioactive carbon distribution in the corrosion zone of IX17 steel at the depth of 0.01 mm are given on Fig. 4 as a function of 10,200, and 500 hours test time.

The most probable fact is that first of all the admixtures of intrusion which are in *d*-solid solution dissolve in lithium. The violation of balance between impoverished -solid solution and chromium carbides causes their partial dissolution.

The behaviour of carbides present in the structure of steel, as well as of sintered carbides was studied.

For this purpose a series of melts, the chemical compo-

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tion of which (according to the calculations) ensured the presence of only one element in the carbide metal was melted. X-ray structural analysis revealed that the samples contained  $\text{Fe}_3\text{C}$ ;  $\text{Cr}_{23}\text{C}_6 + \text{Cr}_7\text{C}_3$ ; NbC carbides.

The tests were carried out in the containers made of IXI3M2 at  $800^\circ\text{C}$  during 25 hours.

Iron carbides are easily destroyed by lithium. Carbon content on the surface of steel decreased from 1.6 to 0.04 per cent, and in lithium increased from  $1.10^{-3}$  per cent to  $1.95 \cdot 10^{-2}$  per cent.

Having been exposed to lithium, the chromium carbides were not destroyed, however they underwent some changes. The weight of a sample increased by 0.07 per cent, while on the internal surface of the container, in which the high chrome sample with C=0.5 per cent and Cr=25 per cent was tested carbon content decreased from 0.1 to 0.06 per cent to the depth up to 0.25 mm. Thus, the carbon was transferred from the surface of the container to the sample with higher chrome content than that of the container. It is possible that at the same time the change of composition of carbides in the sample took place, and some  $\text{Cr}_{23}\text{C}_6$  carbide was transformed into  $\text{Cr}_7\text{C}_3$  carbide.

The changes of chrome carbides were also observed when testing  $\text{Cr}_3\text{C}_2$  sintered carbides in the containers made of armco iron. In the samples tested at  $1100^\circ\text{C}$  the change of structure to the depth up to  $200\mu$  was found. As X-ray structural analysis showed, at the above mentioned depth  $\text{Cr}_3\text{C}_2$  carbide had been transformed into  $\text{Cr}_7\text{C}_3$  carbide. The transfer of chrome on to the container made of armco iron was also observed.

After testing  $\text{Mo}_2\text{C}$  sintered carbide in lithium at  $800^\circ\text{C}$  during 100 hours its weight losses make up about 1 per cent, and at  $1100^\circ\text{C}$  - up to 3 per cent. The X-ray structural analysis of  $\text{Mo}_2\text{C}$  sample tested in lithium at the temperature of  $1100^\circ\text{C}$  showed the presence of thin ( $\sim 5\mu$ )  $(\text{Mo}, \text{Fe})_6$  C-type carbide layer on the surface.

Niobium carbides in the structure of steel are mainly inside the grains. After testing in liquid lithium at  $800^\circ\text{C}$  during 25 hours, their form under microscope remains the same as

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it was before the test. Carbon content in liquid lithium as well as on the surface of the container did not differ .

After the test from the initial one.

NbC sintered carbides are also of high stability in lithium. NbC corrosion was not observed after testing at 800- $1100^{\circ}\text{C}$  during 100 hours.

The behaviour of titanium carbides in lithium is similar to that of biobium carbides.

It was shown <sup>8</sup> that lithium diffusion in armco iron mainly proceeded along the bounds of grains and was in direct connection with the contamination of bounds with the admixtures (carbon, oxygen, sulphur, etc.), with which lithium was of considerable chemical affinity. In the iron and chrome steels melted in vacuum and having considerably less quantity of the above mentioned admixtures the diffusive mobility of lithium sharply decreases.

Iron atoms, as well as chrome atoms are dissolved and transferred into lithium. The vacancies which may be accumulated in the surface layer of some thickness, as well as move into deeper layers appear on the surface of constructional material. Lithium diffuses through the vacancies and their accumulations, therefore its concentration on the surface is higher, and the depth of penetration corresponds to the layer, where the decrease of chrome content was found. However, lithium concentration in the surface layer substantially depends on the composition of steel. For example, lithium content in IXI3 steel is 6-10 times higher than in IXI2M2 БФ steel. Lower lithium content on the surface of IXI2M2 БФ steel may be explained by the fact that the defects of crystalline lattice, which appeared after the dissolution of chrome in lithium, can be healed with molybdenum atoms, the dissolvability of which in lithium is many times less than that of chrome.

The stability of non-metallic phases, such as: oxides and sulfides present in IXI3M2 chrome steel was investigated. The samples containing the oxides were prepared in the following way: before the end of the melting, IXI3M2 steel was usually blown through with oxygen, and some metal was poured out without any deoxidization. The following portions of metal were usually deoxydized separately by ferrosilicon, silicocalcium, aluminium,

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calcium and nickel-magnesium alloy.

The investigations showed that the inclusions of  $\text{Cr}_2\text{O}_3$ ,  $\text{Al}_2\text{O}_3$  and  $\text{SiO}_2$  oxides, as well as  $\text{SiO}_2$ ,  $\text{CaO}$  calcium silicates were usually destroyed in lithium at  $800^\circ\text{C}$ , and on the the hollows usually appeared in place of oxide inclusions. The inclusions of  $\text{MgO}$  and  $\text{CaO}$  oxides do not interact with lithium at  $800^\circ\text{C}$ . The interaction of lithium with oxide inclusions agrees with the data on their thermodynamic stability.

The tests of IXI3M2 steel (in which during its melting the sulfide inclusions of iron, manganese, titanium, aluminium and calcium were obtained) in lithium at  $800^\circ\text{C}$  during 25 hours showed that the sulfides of all the above-mentioned elements were usually destroyed in lithium. Only the sulfides of some rare earth elements (cerium, lanthanum) turned out to be stable in lithium.

Unsatisfactory stability of oxides and sulfides (which are usually present in steels along the bounds of grains) in lithium results in the decrease of corrosion resistance of steels and in the development of intercrystalline corrosion during their interaction with lithium. In order to increase the corrosion resistance of stainless steels in lithium one should pay especial attention to the improvement of metallurgical process of steel production (the application of vacuum melting, accurate deoxidization, etc.).

#### b) Corrosion of stainless Steels in Lithium Flow.

(Tests in Convection Barrels)

The tests in convection barrels were carried out at the temperature of a hot zone  $650^\circ\text{C}$  and that of cold zone  $520^\circ\text{C}$  during 1000 hours. In order to create the best convection the barrels were set at an angle of  $45^\circ$  during the test. Chrome and chrome-nickel stainless steels, such as: IXI3, IXI2M2BФ, IXI3H9T, IX20hI4C2, were investigated. In the presence of a hot zone, where the dissolvability of constructional materials in lithium is higher than that in a cold zone, the concentration gradient of dissolved elements is continuously maintained, owing to their crystallization in a cold zone.

Oxygen content in initial lithium made up  $1 \cdot 10^{-2}$  per cent; higher oxygen content (about 1 per cent) was provided by int-

reducing  $\text{Fe}_2\text{O}_3$  oxides. In all cases nitrogen content made up about  $1.10^{-2}$  per cent. The plate samples of 0.5 mm thick which were inserted into the special socket sleeves (placed into both the hot and cold zones of barrel) underwent the test.

The chrome steels which were tested in the barrels (made of IXI3 steel) in convection conditions were of sufficiently high corrosion resistance (in compliance with the values of weight loss) in lithium containing  $1.10^{-2}$  per cent.  $\text{O}_2$  Corrosion rate is about  $5-6.10^{-3}$  mm per year Austenitic steels have considerably higher corrosion rate of about  $6-8.10^{-1}$  mm per year under the same conditions. In the presence of 1 per cent  $\text{O}_2$  in lithium the corrosion rate of these steels increases 2-3 times.

The corrosion resistance of chrome steels sharply decreases (approximately 70-90 times) when testing them in the barrels made of austenitic steels. After the convection tests in lithium with small oxygen contamination and in the barrels (made of chrome steel) chrome steels practically do not change their mechanical properties. The contamination of lithium with considerable quantity of oxygen also unessentially effect on the mechanical properties. However, when testing the chrome steels in the barrels made of austenitic steel, their strength and ductility catastrophically decrease. Chemical-spectral analysis showed that after these tests lithium which was taken from the cold zone of the barrel contained Cr up to 1.8 per cent, Ni up to 2.5 per cent and Fe up to 1.5 per cent.

There were found no essential changes in the structure of chrome steels when oxygen content in lithium was higher. Only under the conditions when the materials were dissimilar the samples of chrome and chrome-nickel steels an had corrosion interaction to the depth of up to  $125 \mu$  (Fig. 5), and the ferrite-perlite structure of chrome steels became a ferrite-carbide one with the considerable enlargement of grain size up to 1-2 numbers. In the chrome-nickel steels along with the uniform dissolution of surface layers the intensive extraction of secondary phases took place both along the body of grain and along its bounds.

c) Corrosion of stainless Steels in Flowing Lithium  
of (in "Mixer" - type plant). The tests with the use

of "Mixer" - type plant were carried out at 700°C for up to 1000 hours at 6 m/sec. speed of travel of samples relative to lithium. 18-8 and 20-14 austenitic steels in form of cylindrical samples of 3 mm in diameter were investigated.

After the test in lithium there was practically no corrosion zone (which is usually determined metallographic alloy) on all the samples, but the diameter decrease of the samples by 100-300 was observed. The mechanical characteristics remained sufficiently high. The diameter decrease of samples shows that the corrosion layer is of small strength and is destroyed in the lithium flow. The corrosion layer has some defects of crystalline lattice as well as small strength of individual grain bonds and therefore is easily destroyed in lithium flow.

Thus, the nature of destruction of constructional materials in the flow of heat-transfer metals is determined not only by the density of liquid metal and the speed of flow, but also by the strength of bonds between individual volumes of corrosion layer.

The presence of zirconic getter in lithium 500 cm<sup>2</sup> zirconium per 500 cm<sup>3</sup> lithium) positively effects on the corrosion resistance of austenitic steels: with the use of "Mixer" corrosion rate decreases by one order. According to the published data and thermodynamical calculations (9), zirconium is a good getter for nitrogen, and this may account for the increase of corrosion-erosive resistance of austenitic steels in lithium.

#### CONCLUSIONS

1. The chemical nature of oxygen dissolved in liquid sodium (oxygen ions and Na<sub>2</sub>O ed excited molecules) is the determining factor accelerating the corrosion of materials in sodium coolant contaminated by oxygen.
2. At 500 and 600°C the decarbonization of low-alloy perlite steels is observed in sodium flow. If the content of carbide-forming element, i.e. chrome, increases to 5 per cent and higher, the decarbonization of steels stops.
3. The experimental results on corrosion resistance of austenitic steels in 5 m/sec. sodium flow at 600 and 700°C

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Showed that oxygen content in sodium coolant had not to exceed  $5 \cdot 10^{-3}$  per cent.

4. It was shown that the tendency of nickel base refractory alloys to the mass transfer was determined by nickel content. Aluminium additions noticeably decreased this tendency.

5. Molybdenum in sodium coolant showed high corrosion resistance up to  $900^{\circ}\text{C}$  when oxygen content was from  $2 \cdot 10^{-3}$  to  $1 \cdot 10^{-2}$  per cent. The corrosion resistance of niobium to a great extent depends on an oxygen content in sodium.

6. The basic reason for the corrosion of stainless chrome and chrome-nickel steels in lithium is the destruction of non-metallic phases, (carbides, oxides and sulfides) as well as isothermal and non-isothermal transfer of the dissolved atoms of steel components (nickel, chrome, iron, etc.).

7. Chrome stainless steels which are additionally alloyed with molybdenum, niobium and vanadium are more perspective for lithium coolant.

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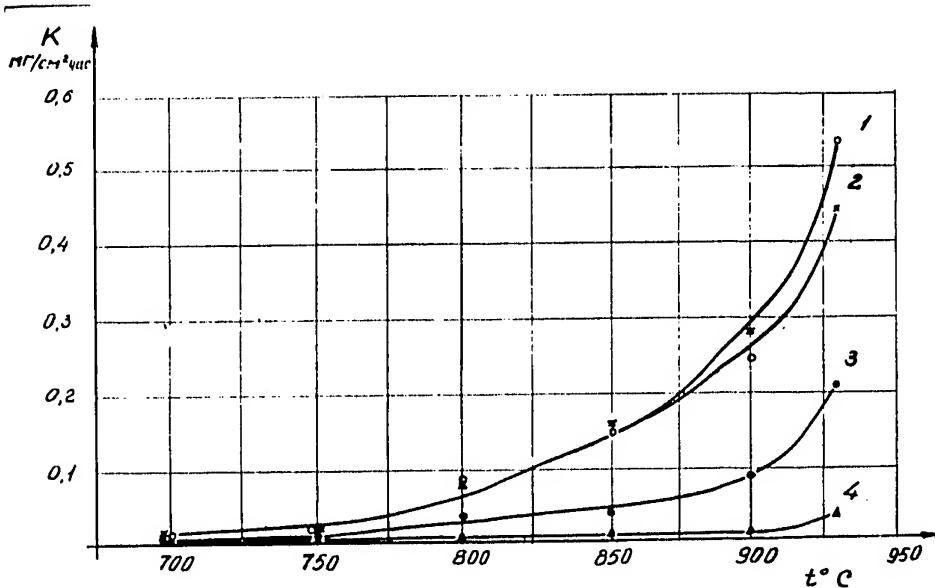


Fig. 1. Tendency of constructional materials to the mass transfer in sodium at 700-930°C.

1 - alloy with Cr=20%; Ni = 75%; 2 - alloy with Cr = 20%; Ni = 60%; Mo = 5%; W = 15%; 3 - Alloy with Cr = 16%; Ni = 60%; Mo = 5%; W = 15%; 3 - alloy with Cr = 16% Ni = 60%; Al = 3%; 4 - Steel with Cr = 20%; Ni = 14%.

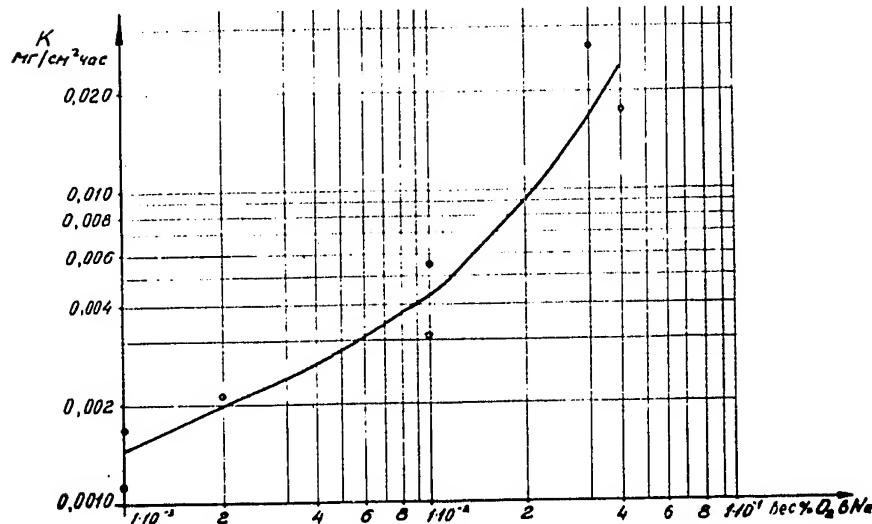


Fig. 2. Dependence of niobium corrosion resistance on oxygen content in sodium at 900°C

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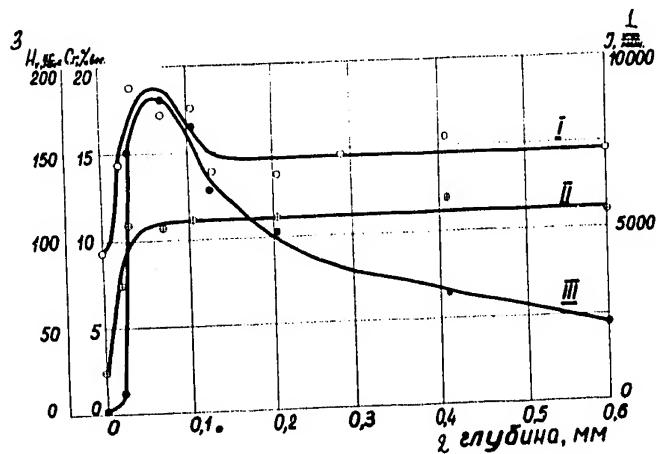


Fig.3(a) Distribution of  $\text{C}^{14}$  isotope (III), chrome (II) and the change of microhardness (I) in OX13 steel tested in the containers made of armco iron at  $800^\circ\text{C}$  during 200 hours  
 1 - count/min; 2 - depth, mm;  ${}^3\text{H}_{\text{kg}} \text{mm}^2$ , Cr, weight per cent

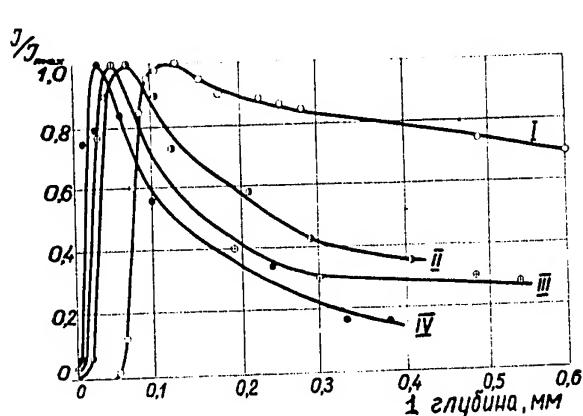


Fig.3(b) Distribution of  $\text{C}^{14}$  in OX13 steel tested in lithium in containers made of armco iron at  $800^\circ\text{C}$  during 500, 200, 50, and 10 hours (I-IV, respectively)  
 1 - depth, mm

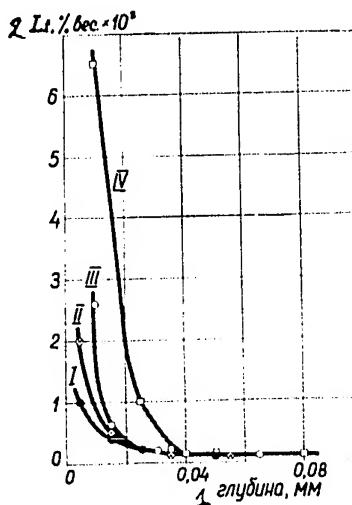
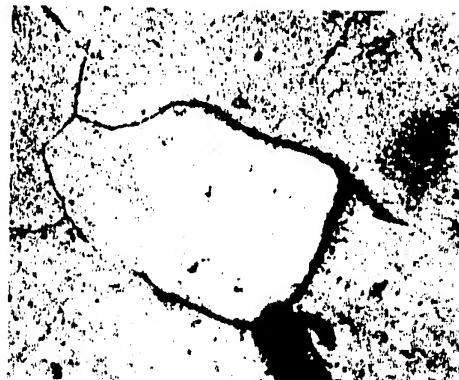


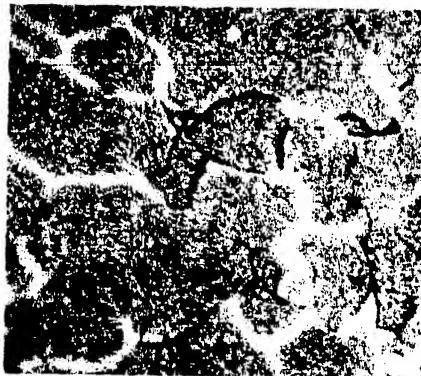
Fig.3(c). Lithium distribution in 1x12 M2 6Φ steel tested in lithium in container made of the same steel at  $700^\circ\text{C}$  during 50, 200, and 500 hours (curves I, II, III, respectively) and at  $800^\circ\text{C}$  during 500 hours (curve IV)  
 1 - depth, mm 2 -  $L_f$ , weight per cent  $\times 10^3$



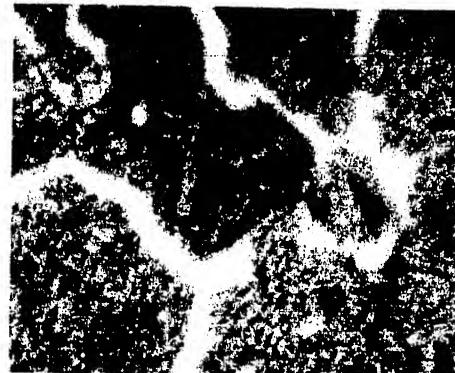
a) 10 hours. Carbides along the grain bounds and inside the grains.  $\times 450$



b) 500 hours. Almost all the grain bounds are dark(destroyed by lithium)  $\times 70$ .



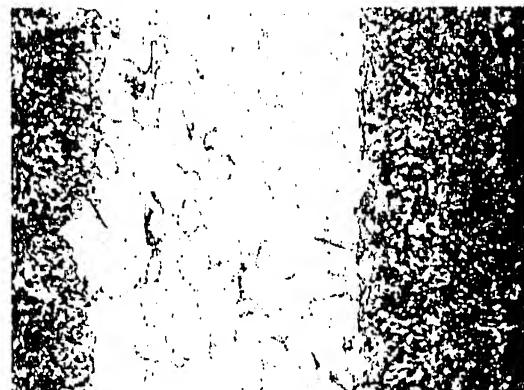
c) 200 hours. Grain bounds are partially decarburized.  $\times 70$



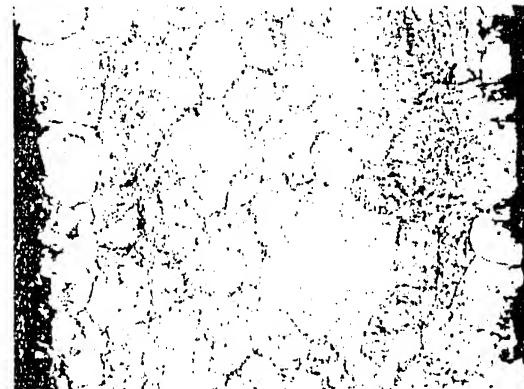
d) 500 hours. Light decarburized grain bounds.  $\times 70$

Fig.4. Micrographs (a,b) and auto-radiographs (c,d) of 1x17 steel at the depth of 0.01 mm after testing it in lithium at 800°C in the containers made of armour iron

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a) 1 x 12M2 BΦ steel tested in lithium in the convection barrel made of 1 x 18 H 9 T steel.



b) 1 x 20 H 14 C 2 steel tested in lithium; convection barrel made of 1 x 18H9T steel.

Fig. 5.

Microstructure of stainless steels after convection testing in lithium at 650°C and at temperature gradient 130°C during 1000 hours. x 200.